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SOME OBSERVATIONS ON THE STRUCTURE OF TI-11.5MO-6ZR-4.5SN (BETA--ETC(U)
MAR 77 J C WILLIAMS, F H FROES, C F YOLTON N00014-76-C-0409

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"Some Observations on the Structure of Ti-11.5Mo-6Zr-4.5Sn (Beta III) as Affected by Processing History"

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SOME OBSERVATIONS ON THE STRUCTURE OF TI-11.5Mo-6Zr-4.5Sn (BETA III) AS AFFECTED BY PROCESSING HISTORY

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There have been a number of articles written on the structure and properties of the metastable beta titanium alloy Ti-11.5Mo-6Zr-4.5Sn (Beta III).[1-11] Several of these have made sound contributions to the understanding of the structure and properties of the alloy in the heat treated conditions studied. However, many have not addressed the questions of optimum processing and heat treatment as viewed from an application's standpoint. Accordingly, the alloy has been variously reported to have poorer tensile ductility, [4,7] lower stress corrosion resistance[5] and poorer toughness than is typical when proper processing conditions are used. Furthermore, in some cases the details of the microstructural analysis appear to be in conflict and in other cases the interpretation given to some of these results warrants closer scrutiny with regard to their correctness and general applicability.[4,6,7] For example, Ganesan, et al[7] suggest that ω-phase precipitates in β-grain boundaries, yet many other studies have shown that ω-phase only is uniformly nucleated whereas α-phase is always the heterogeneous nucleation product.[12,13] Further, the effects that spontaneous relaxation in thin foils have on the apparent microstructure are now well-known and have been extensively documented.[14,15] Despite this, the latent effects of such thin foil artifacts have been described in length[6] even though there appears to be no correlation between these effects and bulk

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material behavior.

The purpose of this communication is to describe the optimum processing for Beta III, to demonstrate the effect of such processing on mechanical properties and to correct several erroneous notions regarding microstructural details of the alloy.

Since Beta III is generically known as a metastable beta alloy, many investigators have been tempted to start with the alloy in the fully metastable β-phase condition. This is achieved by solution treating the alloy above the beta transus followed by rapid cooling to room temperature. Subsequent aging of the alloy in this condition results in precipitation of the ω or α phases, depending on the aging temperature. Numerous investigations have examined the formation of omega phase in this and other alloys and these have shown that the $\beta \rightleftarrows \omega+\beta$ reaction occurs rapidly during aging at temperatures between 315°C (600°F) and 455°C (850°F).[11,12,16] It generally agreed that the formation of large volume fractions of ω-phase leads to large increases in strength and drastic reductions in ductility. Under carefully controlled circumstances it has been shown that attractive properties can be achieved in the $\beta+\omega$ condition.[10,13] However, the rapid kinetics of ω-phase formation make the control of ω-phase volume fraction difficult to the point of impracticality. Thus, means of minimizing or eliminating ω-phase formation are desirable, and under no circumstances can processing to yield omega phase be considered as optimum processing as has been suggested elsewhere.[7] Aging at 480°C (900°F) and above leads to α-phase precipitation and an attendant sizeable increase in strength with the retention of good ductility.[1] Representative properties for

these heat treated conditions are listed in Table I. The kinetics of α-phase formation by uniform nucleation are much slower than those of ω-phase formation.[1] As a result, the tendency for heterogeneous α-phase nucleation is very pronounced.[8] This tendency leads to extensive α-phase formation at β-phase grain boundaries (Figure 1) unless a suitable density of alternate nucleation sites are present. Such nucleation sites include dislocations and dislocation sub-boundaries, the density of which can be controlled by warm working* the material prior to aging. The nucleation of α-phase at sub-boundaries is shown in Figure 2. Under such conditions, nucleation of α-phase at β grain boundaries can be minimized or suppressed.[8] Further, aging warm worked material which contains a high dislocation density can suppress ω-phase during aging. This results from the marked acceleration in kinetics of α-phase precipitation in the presence of heterogeneous nucleation sites. These observations tend to cast doubt on suggestions by other workers that ω-phase can form at grain boundaries.[7] No evidence for heterogeneous nucleation of ω-phase has been obtained in the investigations described herein.

Based on previous results[10] which suggest that ω-phase formation is difficult to control because of the rapid reaction kinetics, we strongly recommend that any optimum processing sequence must result in a final microstructure which does not contain ω-phase. Based on the earlier discussion, this requires warm working to provide sufficient α-phase nucleation sites to ensure that ω-phase formation is suppressed. Further, earlier discussion also showed that optimum processing results in suppression of grain boundary α. This also can be achieved by warm working since a

* We use "warm working" to describe working above room temperature but below the recrystallization temperature.

high density of intragranular nucleation α-phase sites suppresses grain boundary α-phase precipitation. Thus, our suggested optimum processing of β-III is a warm working operation, for example a 50% reduction in the temperature range 730°- 675°C, followed by an aging treatment the duration and temperature of which is selected to give the desired strength level. In this discussion we have considered that current increasing emphasis on fracture control in structural number places a practical upper strength limit in the neighborhood of 1240-1275 MPA (180-185 ksi) ultimate tensile strength and 1170-1210 MPA (170-175 ksi) yield strength. Such treatments as pre-aging in ω-phase formation followed by a higher temperature aging treatment[4] lead to much higher strengths but these have very limited interest for structural applications. In this context, we suggest that this optimum processing provides improvements in toughness: strength, stress corrosion resistance and tensile ductility at any particular strength level when compared to non-optimum processed material. An example of this latter might be material which has been super transus solution treated, quenched and aged.

The uniformity of α-phase heterogeneous nuclation sites (dislocations) depends to a significant extent on the deformation mode; planar slip or twinning are undesirable in this regard since they result in inhomogeneous deformation. It has been reported that Beta III exhibits a grain size dependent twin-slip transition during room temperature deformation.[4] We have examined this point and can find no evidence for such a transition. Samples with grain sizes ranging from 6μm to 95μm were deformed ∿10% at room temperature and examined by transmission electron microscopy. In all cases twinning

was observed, an example of which is shown in Figure 3. Both electron diffraction and x-ray diffraction were used to verify that the lenticular deformation product shown above was twinning and not a strain-induced martensite as has been suggested and discussed elsewhere.[7,17] Only bcc reflections were obtained in the diffraction patterns which verified the product as twinning rather than martensite.

In summary, we have shown that warm working of Beta III so as to promote a high residual dislocation density has a marked influence on microstructure in the following regard.

1. The presence of a high dislocation density promotes transgranular nucleation of α-phase and accelerates the kinetics of α-phase formation. Both of these factors tend to suppress formation of detrimental grain boundary alpha.
2. The presence of a high dislocation density promotes direct formation of the equilibrium α-phase and thus suppresses formation of the undesirable transitional ω-phase.
3. Deformation of Beta III at room temperature always results in twinning whereas elevated temperature deformation can occur by slip alone.

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LIST OF FIGURE CAPTIONS

Figure 1. Bright field electron micrograph of Beta III solution treated above the beta transus and aged 8 h at 950^{o}F (515^{o}C), showing continuous layer of alpha phase at beta grain boundary and fine uniformly nucleated alpha phase within the beta grains.

Figure 2. Bright field electron micrograph showing heterogeneously nucleated alpha phase precipitates at dislocation boundaries.

Figure 3. Bright field electron micrograph showing twins in Beta III solution treated above the beta transus and deformed $\sim$10% by cold rolling.

TABLE I

TYPICAL AGED PROPERTIES OF BETA III BAR, PLATE AND SHEET

Product	Condition	Tensile Strength (ksi)	Yield Strength 0.2% Offset (ksi)	Elongation (%)	Reduction of Area (%)
1" to 1½" dia. Bar	Sub-transus ST + 900F 8 hr	194	184	12	36
" " "	" " " 1000F 8 hr	172	164	14	52
1" Plate	Sub-transus ST + 900F 8 hr	190	177	3	11
" "	" " " 1100F 8 hr	151	144	8	18
0.063" Sheet	Sub-transus ST + 950F 8 hr	190	178	7	25
" "	" " " 1000F 8 hr	168	158	8	45

1 ksi = 6.89 MPa
$^{\circ}F = 1.8 \times {}^{\circ}C + 32$
1 in. = 25.4 mm

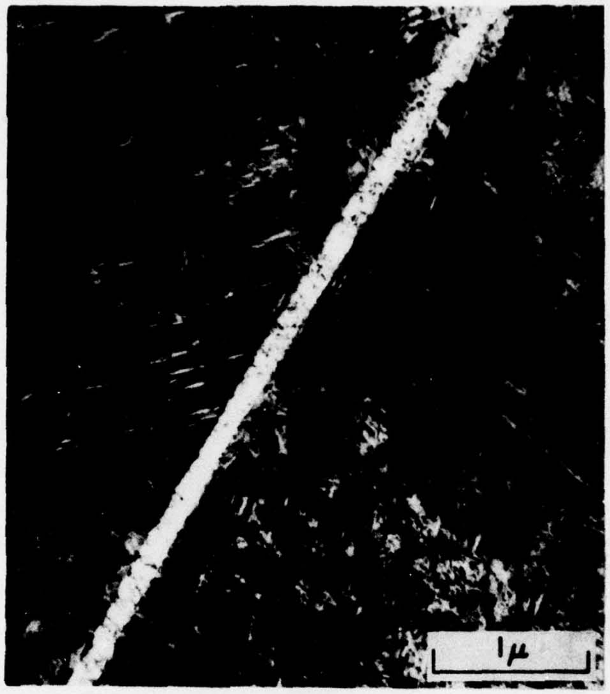

Figure 1

Figure 2

Figure 3

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13. ABSTRACT

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This paper describes the effect of processing history on properties of the metastable β titanium alloy Ti-11.5Mo-6Zr-4.5Sn. It is shown that numerous earlier published accounts of properties have corresponded to non-optimum processing. In addition, attempts to correlate microstructure with properties have often been based on questionable microstructural interpretation; this paper also attempts to correct several of these points. Finally, the paper describes what is considered to be closer to if not optimum processing of this alloy.

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